PUBLIC LIBRARY
MAY 11 1938
DETROIT

SUPPLEMENT TO THE ENGINEER. APRIL

The Metallurgist.

Supplement to The Engineer, *April, 1938.*

Thermodynamics and Metallurgy.

THE early study of chemical reactions was directed almost exclusively to the material changes which took place, the nature of the substances before reaction, the velocity and extent of the reaction, and the nature of the products. Attention began to be directed, about the middle of the last century, to the quantitative study of the thermal changes which accompanied the reactions as a means of throwing light on the problems of chemical "affinity." With the development of the science of thermodynamics, more exact means of determining the motive forces of chemical reactions and the governing conditions of chemical equilibrium have become available. The chemical reactions underlying the practice of metallurgy are markedly influenced by temperature conditions. By utilising the thermodynamic relationship between temperature, heat of reaction, and the equilibrium constant our knowledge of the reactions involved in oxidation, and reduction in desulphurisation and decarburisation, and in the smelting and refining of metals generally can be established on a firmer theoretical basis. One thing is lacking for this purpose, and that is the necessary accurate information about heats of reaction, specific heats, latent heats of fusion, &c., at the very high temperatures involved. For such work the metallurgist has hitherto been almost wholly dependent on isolated investigations of chemists and physicists.

The collection and correlation of thermodynamic data concerning substances of metallurgical interest is therefore a matter of great importance, and the issue by the United States Bureau of Mines of the eighth bulletin* in the series entitled "Contributions to the Data on Theoretical Metallurgy," is to be welcomed. This paper deals with "The Thermodynamic Properties of Metal Carbides and Nitrides." These data are difficult to obtain experimentally and are consequently rare, and possibly in some cases not very accurate. The work of Mr. Kelley in collecting them from so wide a field, and in critically assessing their relative importance, is for this reason, the more valuable. The first part of the paper deals with the carbides in alphabetical order, and the second part with the nitrides. The data given include specific heats, heat and free energy of formation, and heats of reactions in which the compound is involved. The paper concludes with an interesting discussion of the application of the data to half a dozen problems of practical interest. These include the carburisation of iron by methane, the debismuthising of lead by calcium through the medium of calcium carbide, and the partition of carbon between iron and manganese in the γ-range. Of special interest is the section on the decarburisation of chromium. Ferro-chromium made by carbon reduction of chromite ordinarily contains 4 to 8 per cent. of carbon. The carbon content of this product may be reduced in several ways, one industrial method being a treatment with a fused mixture of chromite, lime, and fluorspar. The possibilities of decarburising ferro-chromium with chromite, with silicon, or with calcium, are discussed. The exact formula of the chromium carbide present in the ferro-chromium is not known with certainty, but it is shown that it makes little practical difference whether the carbon is assumed to be combined as Cr_3C_2 or Cr_5C_2. Calculations are made for both possibilities.

From the equilibrium constant of the reaction :

$$3\,Cr_3C_2 + 2\,Cr_2O_3 = 13\,Cr + 6\,CO,$$

calculated for various temperatures it is deduced that carbon removal is not effective below 1800 deg. Abs. and that a temperature of 2000 deg. Abs. is required for obtaining a low carbon content. The reaction involving the other main constituent of chromite, FeO, would be more effective than the one under consideration.

From the value of the equilibrium constant for the reaction

$$Cr_3C_2 + 2\,Si = 3\,Cr + 2\,SiC$$

it appears possible at temperatures above 1700 deg. Abs. to decarburise by silicon and produce a low carbon product, provided that equilibrium is reached and enough silicon is added. To reduce carbon from 6 per cent. to 0·1 per cent. in 65 per cent. ferro-chromium by means of silicon would result in a product containing at least 1·5 per cent. of silicon. Further deductions are that considerable removal of silicon from this product, without serious loss of chromium, may be possible by mild oxidation, although direct oxidation with oxygen is not practicable, that silicon carbide SiC is reducing to both Cr_2O_3 and FeO at a sufficiently high temperature, and that a process might be realised in which chromite is reduced by carbon to form a high-carbon ferro-chromium, which in turn is treated with silicon to lower the carbon and most of the low-carbon ferro-chromium tapped off, after which smelting might be continued, use being made of the SiC formed as part of the reducing material for the following batch of chromite.

Decarburising by means of calcium by making use of the reaction

$$Cr_3C_2 + Ca = 3\,Cr + CaC_2$$

does not appear to be feasible. The product would contain 2·2 per cent. of calcium when the carbon was down to 1 per cent. and 18·5 per cent. (if such a concentration could be obtained in a chromium-iron

* K. K. Kelley, U.S. Bureau of Mines, Washington, D.C., Bulletin 407, 1937, 66 pp.; price 10c.

alloy) when the carbon was reduced to 0·1 per cent. If the calcium were present as a separate liquid phase the decarburisation would be dependent on diffusion and would necessarily be slow; moreover, at the lowest possible decarburising temperature the vapour pressure of calcium closely approaches one atmosphere and oxidation of calcium would be a serious matter. This excessive loss of calcium in attempts at carbon removal has been confirmed in experiments carried out in the Metallurgical Division of the Bureau of Mines.

A glance through the Bibliography appended to the Bulletin shows that very little of the information is derived from work carried out in this country. In addition to the late Dr. Bramley, the British investigators mentioned are Bone and Jerdan, Coward and Wilson, Hutton, Pring and Slade clearly showing the stimulating influence, on this work, of the Manchester school of research on the phenomena of combustion. The main contributions made by British investigators to the physical chemistry of metallurgical reactions have originated under this influence and reached their greatest metallurgical interest in the study of blast-furnace reactions carried out under the direction of Professor Bone of the Imperial College, South Kensington. The investigation of the heat of formation of solid compounds of metallurgical importance, their specific heats at very high temperatures, and the influence of temperature on their thermal characteristics is apparently more highly developed in Germany and America.

A general discussion on "The Physical Chemistry of Steel Making Processes," held by the Faraday Society in 1925, served to emphasise the need of more information on these lines. The position at the time was admirably summarised by Dr. McCance, and for this reason many will look forward with interest to the paper on "The Application of Physical Chemistry to Steel Manufacture" which is to be given by Dr. McCance at the May meeting of the Iron and Steel Institute. Though much necessary information may still be lacking, an indication of the scarcity of thermodynamic data in directions in which they are urgently needed is a necessary prelude to the further attack on the problems by experimental methods which, it may be hoped, will be stimulated by the forthcoming discussion.

Identification and Sorting of Light Metal Scrap.

By EDMUND RICHARD THEWS.

WHILE the use of scrap metals in connection with the production of ferrous and non-ferrous alloys has formed an important economic requirement for many years, it has gradually been realised that in many instances the addition of suitable proportions of well-selected scrap metals to new metal mixtures actually introduces marked technical advantages as well. Indeed, it has been found that in the case of many alloys, including complex copper alloys and many aluminium alloys, the best results can only be obtained if, apart from the remelting treatments frequently required, the mixtures contain from 20 to 40 per cent. of suitable clean scrap.

It is quite obvious, of course, that satisfactory results can only be obtained if the scrap employed is perfectly adapted to the purposes in question, *i.e.*, if it has been sorted with due attention to composition and freedom from impurities. In most cases, the foundries and melting works employ their own scrap exclusively for purposes requiring absolute accuracy of composition and freedom from deleterious elements and impurities, but most of the general scrap used in the metal industries must of necessity be procured in the open market.

Sorting and classification of scrap may, or may not, be difficult, according to circumstances.

It is absolutely impossible, of course, even for the most expert sorter to determine the exact quantitative composition of scrap metals, and chemical analysis should always be applied before the scrap metal is added to alloy mixtures. Not every piece of scrap can be separately analysed, however, and the metal industries, therefore, rely on the sorter to grade the scrap into lots which can then be melted together and analysed as a whole.

The identification and sorting of light metal scrap is much more difficult for the uninitiated than that of copper alloys, or of ordinary iron and steel (excluding that containing a large amount of the special elements). This is due partly to the fact that the aluminium and magnesium alloys as a whole are much more subdivided into chief and side groups than the industrial copper alloys, while the range of composition within these numerous groups is much wider as a rule and much less clearly defined. Furthermore, there is practically no difference of appearance between all the light metal alloys after a certain period of usage. The type of fracture also is of little avail as apart from the comparatively small difference between the various structures, these are greatly influenced by the mechanical and ageing treatments to which they are subjected. It is practically impossible, therefore, without comprehensive metallurgical knowledge of all these facts and without knowing the history of the scrap in hand, to identify it reliably.

Another feature rendering the identification of unknown scrap of this type more difficult than that of heavy metal scrap is that the apparent mode of application as indicated by the shape of the scrapped materials cannot safely be employed to facilitate identification as in the case of the copper alloys. While, for instance, a brass sheet or a brass rod must of necessity be of a composition ranging within comparatively narrow margins, light metal plates, &c., may be of any number of different compositions, some of which are distinctly incompatible with the others if mixed in the remelted state and must therefore be kept separate. Light metal pistons, also, may be composed of widely varying mixtures of the aluminium or magnesium alloy type; while the magnesium alloy pistons are of rather uniform composition, those of aluminium alloys may contain alternatively copper, silicon, nickel, or magnesium as chief components besides aluminium. Thus, while magnesium alloy piston scrap (which usually contains, besides magnesium, only a few per cent. of zinc and silicon) can be collected into one lot, the aluminium alloy piston scrap must be more carefully examined. The same applies to the die cast alloys and to most other industrial groups of light metal alloys.

It is thus evident that intelligent identification and sorting of light metal scrap should be based exclusively on chemical short tests such as are represented by simple drop tests. This method of testing light metal alloys is based on the fact that if a drop of definitely composed solutions is placed

on the cleaned surface of the metal, it will give certain reactions indicating the presence of appreciable percentages of certain elements in the light metal tested. The test is exceedingly simple and rapid, and as most of the scrap handled by the average collectors and foundries, or smelters, is well known, the unknown light metal scrap assuming a secondary position only, its application is quite economical—especially if it is carried out as a routine method.

Before studying the various tests possible by the dropping method, it may be advisable shortly to review the various chief groups of light metal alloys with a view to their mutual influences. Below is a diagram developed by Dr. R. Irmann indicating the limits of miscibility of the chief groups of alloys to be considered.

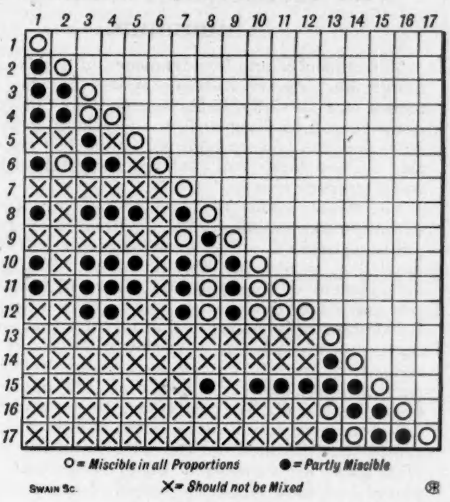
The diagram indicates, for instance, that the groups aluminium-magnesium, aluminium-manganese, and aluminium-magnesium-manganese must not be mixed with the alloys containing copper. The same applies to the aluminium-magnesium-silicon alloys. The latter alloys as well as the three groups above mentioned are characterised by a high degree of corrosion resistance which would be spoiled by mixing with the copper alloys. The difference between the technical properties of the aluminium-magnesium alloys and the aluminium-magnesium-manganese alloys, on the other hand, is not so considerable, and a certain amount of the latter alloys can even be added to the former, as the aluminium-magnesium alloys go fairly well with a limited percentage of manganese.

The aluminium-magnesium alloys and the aluminium-magnesium alloys just mentioned must under no condition be mixed with the silicon-rich aluminium alloys of the Alpax or Silumin type, as this would result in the production of excessively brittle alloys. The same applies to a mixture of the above two groups of alloys with the aluminium-magnesium-silicon alloys, although a small addition of the latter would not cause as considerable an increase of brittleness as the straight aluminium-silicon alloys, of course. Alloys miscible in all proportions are the aluminium-silicon-magnesium alloys, the aluminium-silicon alloys and the aluminium-silicon-copper alloys, but the resulting alloys can only be employed for casting purposes.

It is thus evident that all the alloys indicated by circles within the squares can be mixed in all proportions with the alloys shown at right angles, the restrictedly miscible alloys (not to be mixed if possible) are indicated by black dots in the squares of the intersections, while those absolutely immiscible are indicated by crosses.

It is thus clear that the metallurgist, founder, or scrap collector should always make certain of the general composition of unknown light metal scrap by resorting to the drop test above mentioned.

The drop test, while extremely simple in the hands of metallurgists, chemists, or trained foundrymen, can even be rendered reliable and simple in the hands of untrained sorters if their lack of technical training is compensated by a methodical and logical organisation of the test. The four solutions used for this purpose are kept in individual bottles marked (for use by the untrained man) with numbers or colours, and the sorter is instructed accordingly.



1. Aluminium-copper-magnesium.
2. Aluminium-copper-nickel.
3. Aluminium-copper.
4. Aluminium-copper castings.
5. Aluminium-zinc-copper castings.
6. Aluminium-copper-nickel castings.
7. Aluminium-magnesium-silicon.
8. Aluminium-silicon.
9. Aluminium-magnesium-silicon castings.
10. Aluminium-silicon castings.
11. Aluminium-silicon-copper castings.
12. Aluminium-silicon-magnesium castings.
13. Aluminium-magnesium.
14. Aluminium-magnesium-manganese.
15. Aluminium-manganese.
16. Aluminium-magnesium castings.
17. Aluminium-magnesium-manganese castings.

Miscibility of Chief Groups of Aluminium Alloys.

The composition of these four solutions is as follows :—

Solution No. 1.—30 per cent. nitric acid solution in water.

Solution No. 2.—20 per cent. caustic soda solution (20 gr. of caustic soda sticks in 80 c.c. of water).

Solution No. 3.—5 per cent. hydrochloric acid solution in water.

Solution No. 4.—5 per cent. cadmium sulphate solution (5 gr. cadmium sulphate in 95 c.c. water).

These four solutions suffice to identify most of the light metal alloys likely to enter the open market in form of scrap.

The table on page 116 gives the different colorations produced by the four solutions above described.

The solution No. 1 differs from the other three solutions in that it yields a positive reaction only with magnesium alloys of the Electron or Dow metal types. If a perfectly clean spot on the surface

of such metal is covered with a drop of solution No. 1 it will produce a pronounced white coloration after about five minutes. This reaction, which is not obtained with any other group of white metals, is accompanied by a distinct development of gas noticeable through the formation of tiny bubbles within the drop. This test suffices for the identification of scrap rich in magnesium. However, if it is desired to render the identification still more reliable, it is necessary only to place a drop of solution No. 2 on another clean spot of the scrap. If, after ten minutes of reaction, this drop does not effect any visible coloration, magnesium-rich alloys are indicated beyond any doubt, as all alloys rich in aluminium (including those containing smaller percentages of magnesium) will produce a clearly visible reaction. An experienced metallurgist or foundryman will find another confirmation in the relatively light weight of the Electron or Dow metals (specific weight = 1·8; specific weight of Duralumin = 2·8).

Solutions for Drop Tests.

	Solution No.			
	1	2	3	4
Pure aluminium	None	White	—	None
Aluminium-copper...	None	Black	Black remains	—
Aluminium-copper-nickel ...	None	Black	Black remains	—
Aluminium-copper-zinc... ...	None	Black	Black removed	Grey
Aluminium-silicon	None	Greyish brown	Colour remains	—
Aluminium-silicon-copper ...	None	Grey	Colour remains	—
Aluminium - silicon - magnesium	None	Greyish brown	Colour remains	—
Aluminium-magnesium... ...	None	White	—	Grey
Aluminium-magnesium-silicon	None	Greyish brown	Colour removed	—
Aluminium - magnesium - manganese	None	White	—	Grey
Magnesium	White	None	—	—

If, on testing the scrap with a drop of the solution No. 2 a white coloration is obtained, the piece may be (1) pure aluminium, (2) aluminium-magnesium, or (3) aluminium-magnesium-manganese. Pure aluminium is indicated if no etching reaction is obtained with the solutions Nos. 1, 3, and 4.

The second group of alloys giving a white stain with the solution No. 2 (aluminium-magnesium) is characterised by a grey coloration caused by a drop of the solution No. 4 after about eight to ten minutes of reaction.

If a drop of the solution No. 2 produces a black stain, the alloys containing copper (aluminium-copper, aluminium-copper-zinc, and aluminium-copper-nickel) are indicated. If on wiping off the remaining caustic soda solution a drop of the solution No. 3 removes the black stain, the scrap tested belongs to the group aluminium-copper-zinc, which is further distinguished from the other two groups of aluminium-copper alloys by a grey stain caused by a drop of solution No. 4. The two groups aluminium-copper and aluminium-copper-nickel cannot be distinguished from one another by the drop test, but mixing of these two groups does not produce any harm outside of diluting the nickel contents.

If a drop of solution No. 2 produces a pure grey stain on the clean metallic surface, the group aluminium-silicon-copper is indicated, the grey spot remaining if (after removing the caustic solution) it is covered with a drop of solution No. 3.

The alloys aluminium-silicon, aluminium-silicon-magnesium, and aluminium-magnesium-silicon are indicated by a greyish brown coloration under a drop of solution No. 2. The latter alloy is distinguished from the other two by the removal of this stain under a drop of solution No. 3 (after removal of the caustic solution No. 2).

In many instances the purchaser of the scrap is informed as to the general composition of the material offered him, and the above comments need be applied only to determine off-hand whether this information is correct. It is obvious, of course, that before buying a considerable lot of unknown scrap metal, the exact composition must be determined by chemical analysis.

The Electrolytic Polishing of Aluminium.

THE difficulty and inconvenience of polishing soft metals by ordinary methods is well known. Such metals are specially liable to take up particles of abrasive and to show surface alteration due to flow. In aluminium of very high purity (99·998 per cent.) this surface layer will undergo recrystallisation at room temperatures.

For these reasons P. Jacquet* has suggested the extension to aluminium of the method of electrolytic polishing devised by him some years ago, and already applied to various metals and alloys. The general arrangement of the bath and electrical circuit has previously been described.† The electrolyte employed for aluminium is similar to that used for lead and tin, and is made by pouring, very slowly, 785 c.c. of acetic anhydride into 215 c.c. of an aqueous solution of perchloric acid (density 1·480), the liquid continuously being kept cool. The specimen to be polished is immersed vertically in the solution at a position some centimetres distant from the cathode, which consists of a plate of aluminium. The electrolytic cell is placed in a bath of flowing water. Direct current is used at a pressure of 50–100 volts. The circuit also contains a variable resistance and an ammeter. The current first rises quickly, attains a maximum value and then decreases. The resistance is adjusted to give a current of 3–5 amperes per square decimetre of anode surface. During electrolysis the solution may be agitated, and its temperature ought not to exceed 45–50 deg. Cent. The time of treatment depends on the initial state of the surface. For example, a sheet of pure aluminium rolled under good conditions attains a sufficient polish and brilliance in fifteen minutes to be examined at magnifications of 100–2500 diameters. When the degree of polish is considered suitable the piece is removed from the electrolyte, washed carefully in running water, then in distilled water, and finally dried in alcohol and ether. As rapid solution occurs at the liquid surface the specimen should be wholly immersed, electrical contact being made by an aluminium wire fixed to the specimen. If the specimen is a sheet of which only a part need be polished, the position corresponding to the surface level of the liquid is "stopped-off" with a suitable lacquer and the current density reduced to 1 or 2 amperes per square decimetre. The freshly

* *Comptes Rendus*, December 13th, 1937, 205, 1232.
† See THE METALLURGIST, August, 1936, 10, 150.

prepared polishing solution does not give the best results. It should first be electrolysed for a time sufficient to allow 4 or 5 grammes of aluminium per litre to enter the solution.

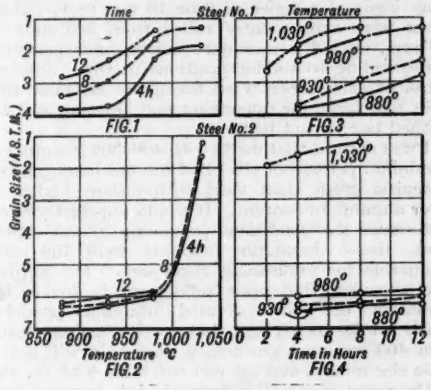
The electrolytic method of polishing can be applied to large articles, or to very thin sheet (for example, 0·1 mm. thick). It is specially suitable for specimens for micro-examination, though if the surface of the specimen has received mechanical treatment the electrolysis should be sufficiently prolonged to dissolve the altered surface layer. Specimens for micro-examination must be examined immediately or kept in a desiccator, sheltered from dust and contamination. After electrolytic polishing the grain boundaries are scarcely visible, but they may be revealed by electrolytic etching in the same polishing solution with a current density of 0·2 to 0·8 amperes per square decimetre. A very short attack reveals the grain boundaries, and a longer one darkens the grains preferentially according to their orientation. As an alternative the usual etching agents may be employed to reveal the structure of the polished surface.

The less pure metal (99·8 per cent.) may be polished with a lower current density than is required for the highly pure material (99·95 per cent. and above). Applied to the macro-examination of ingots electrolytic polishing assists in the observation of physical defects such as cavities and blow-holes, and also reveals the distribution of impurities such as iron and silicon, since the rates of solution of the constituent associated with them are different from that of aluminium. The absence from the polished surface of mechanical deformation or foreign impurities makes it advantageous to use specimens prepared in this way for special work in connection with X-ray and electron diffraction, reflection of light, corrosion, &c., in which the condition of the surface is very important.

Grain Size of Steel.

THE increased importance attached to inherent grain size in steel has led to the publication of papers dealing with a variety of aspects of the subject, *e.g.*, the character of the test, the effect of inherent grain size on response to heat treatment, mechanical properties and machinability, and the practical control of grain size. Among those who have been active in all these directions is the Polish metallurgist, Professor Iwan Fetschenko-Tschopiwski, of the Academy of Mines, Cracow. In conjunction with A. Kalinski,[1] he published a survey of the subject in which much of the American work was repeated on various heats of steel made in a 20-ton basic open-hearth furnace. Many of the properties associated with coarse and fine grain were confirmed. On quench-hardening, the fine-grained steels were superior to the coarse-grained, in that they were less sensitive to quenching temperature and to period of heating and had less tendency to form hardening cracks. The fine-grained steels did not harden deeply, and for a given treatment their maximum hardness was somewhat lower than that of the coarse-grained steels. The coarse-grained steels machined better, but the fine-grained were more wear resistant, and were less brittle after cold working.

The widespread recognition of the significance of austenitic grain size as a factor in the quality of steel has been accompanied by criticisms of the standard McQuaid-Ehn test, and the suggestion of easier and quicker methods of carrying it out. Some such alternative methods have been examined by I. Fetschenko-Tschopiwski and A. Stanislawski,[2] who have studied the behaviour of the austenite grains at different carburising temperatures and different times of exposure to different media. The experiments were carried out on two 0·35 per cent. carbon steels, containing about 0·65 per cent. of manganese and 0·25 per cent. of nickel. The first was made without aluminium additions; to the second 450 grammes of aluminium per ton were added in the ladle. The data derived from tests carried out in a carburising mixture of 60 per cent. charcoal and 40 per cent. barium carbonate are represented in Figs. 1 to 4. In the steel cast without aluminium, grain growth was most rapid between 930 deg. and 990 deg. Cent., though increased carburising time tended to efface the sudden increase. In the steel treated with aluminium there was only a very small increase in



FIGS. 1-4.—*Relation between Grain Size and Time and Temperature of Carburisation in 60 : 40 Charcoal-Barium Carbonate Mixture.* Steel 1, aluminium, nil; steel 2, aluminium, 450 grammes per ton.

grain size until the carburising temperature exceeded 980 deg. Cent., and then an additional 50 deg. produced a very marked increase. In this case the increase was independent of the time of carburising. There is thus a critical temperature below which the steel is immune to grain growth, and above which the grain size is as large as in the aluminium-free steel.

The results of carburising in coal gas were irregular, and no clear difference in structure of the two steels was to be observed. This was probably due to the formation, on the surface, of a 2 mm. thick hard carbonaceous layer derived from the decomposition of hydrocarbons. At a certain stage in its formation further carburisation was hindered, but the carbon already absorbed continued to diffuse inwards. Where the carbonaceous layer was not intact, carburisation proceeded, giving rise to a certain irregularity in the results observed. When the coal gas was saturated with benzol carburisation proceeded a little further before it was checked by the formation of the carbonaceous surface layer, but the results were essentially the same as with coal gas. Special mixtures of charcoal, impregnated with sodium carbonate or cyanide, were also used, but their carburising action was inferior to that of the charcoal-barium carbonate mixture.

These tests show that the rate of decomposition of

the carburising medium must not exceed the rate of diffusion of carbon in the steel, if formation of a carbonaceous surface layer, with consequent irregularity of carburisation, is to be avoided. The advantage of the charcoal-barium carbonate mixture was confirmed, but the authors were of opinion that their results indicated that 900 deg. Cent. was a better carburising temperature than the standard temperature (927 deg. Cent.), and that four hours' carburising was enough to obtain a sufficiently deep cementite network for correct evaluation of grain size.

Another side of the question has been taken up by I. Fetschenko-Tschopiwski and T. Palmrich,[3] in a paper on "The Influence of Time of Casting and Aluminium Addition on the Grain Size of Steel." This is a study of a basic open-hearth carbon steel (carbon, 0·45 per cent.), made in a 40-ton furnace by the method of C. H. Herty, jun., and deoxidised with calcium silicide. Aluminium was added to the ladle and account was taken of the time that the steel remained in the ladle before each successive 4200-kilog. ingot of a series of 9 or 10 was cast. These ingots were subsequently rolled from 500 mm. to 175 mm. or 152 mm. square billets, and specimens corresponding to similar positions in three different ingots of each series were taken for McQuaid-Ehn grain size tests, for quenching and fracture and for notched bar impact tests.

It was noticed that melts with 400–300 grammes of aluminium per ton of steel had a more homogeneous austenitic grain than steel with either higher or lower aluminium content. It would appear therefore that under the conditions prevailing in the experiment, these aluminium amounts were the most favourable for influencing grain size. The authors hold the view that grain refinement is due to the formation of finely divided alumina particles, which act as nuclei, and that with amounts greater than 400 grammes per ton, a part only will act as grain size refiner, and the rest will alloy with the iron in the usual way. The amount, 400 grammes, has no general significance. The actual amount of aluminium required to obtain a definite degree of grain fineness would be expected to depend on the amount of oxygen remaining in the steel, in accordance with the relationship found by Houdremont and Hengstenberg. Comparison of the results of McQuaid-Ehn tests on specimens derived from ingots 1, 5, and 9 or 10 of the series showed that the inherent grain size of the steels treated with aluminium was dependent on the time during which they were in the ladle before casting. Owing to a reaction with the ladle lining as a result of which the alumina particles tend to cohere, thus reducing the number of nuclei, the grain size increased with time in the ladle, and the temperature at which a sudden increase in grain size occurred was lowered. Melts with 350 to 400 grammes of aluminium per ton showed the smallest difference between the first and last ingot. The impact figure of specimens treated to give a constant tensile strength,* showed a tendency to fall from the first to the last ingot. The maximum values, and the least difference between successive ingots, were shown by the melt treated with 300 grammes of aluminium per ton.

Tests were made to determine the depth of hardening of the quenched steels. Specimens, 20 mm. square, were quenched in water from 800 deg., 840 deg., 880 deg., and 920 deg. Cent., and were then fractured. The depth of hardening shown diagrammatically in Fig. 5 was smallest with high aluminium content, with low quenching temperature and with short time in the ladle, *i.e.*, it increased on passing from ingot 1 to ingot 9.

Fetschenko-Tschopiwski and his collaborators may

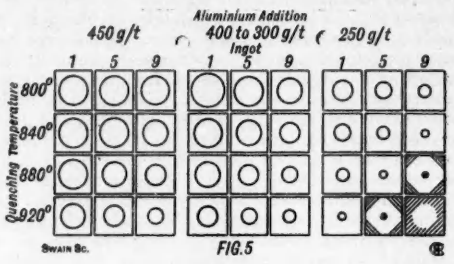


FIG. 5—*Diagram of Fractures of Quenched Specimens showing Extent of Hardened Region and Unhardened Core in relation to Aluminium Addition and Time in Ladle.*

be said to support the views expressed by Dr. Houdremont[4] in the discussion of Swinden and Bolsover's paper as to the necessity of widening the range of the McQuaid-Ehn test to give a space diagram of grain size as a function of heating temperature and time. They reinforce his warning of the necessity of using a standard carburising medium, but they do not advocate multiplying the number of tests. On the other hand, they would reduce the time and the temperature employed in a single carburising test, and combine it with quenching tests. They found that the results of McQuaid-Ehn tests were consistent with, and supplementary to, those of fracture tests on the quenched steels, so that by the use of the two together they were able to make a valuable exploration of the effect on the final product of different aluminium additions and of the changes subsequently occurring in the ladle.

REFERENCES.

[1] *Stahl und Eisen*, 1937, 57, 355; *Journal*, Iron and Stee Inst., 1937 (2), 136, 94A.
[2] *Archiv für das Eisenhüttenwesen*, 1937–38, 11, 287.
[3] *Archiv für das Eisenhüttenwesen*, 1937–38, 11, 405.
[4] *Journal*, Iron and Steel Inst., 1936 (2), 134, 487P; also *Stahl und Eisen*, 1936, 56, 1412, and 1937, 57, 302.

* Stated to be 100 kilos. per square millimetre or 63·5 tons per square inch, which seems very high for a 0·45 per cent. carbon steel, in conjunction with a Mesnager impact figure of over 14 mkg. per square centimetre, *i.e.*, an Izod impact figure probably exceeding 60 foot-pounds.

The Solidification of Metals.

THE mechanism of the solidification of metals is not only of great academic interest, but is also of considerable importance in the foundry. It is well understood that solidification begins by growth from nuclei, whether these are formed in the interior of the melt, giving rise to a granular structure, or at the sides or bottom of the mould, resulting in columnar crystals. Apart from a paper by A. Lange,[1] published in 1931, no measurements of the rate of crystallisation and nucleus formation in metals have, until recently, been published, although some work of this nature has been done on transparent organic materials.

Of great interest, therefore, is a paper recently published by E. Scheil,[2] which records an attempt to measure the nucleus formation experimentally and to account for the results obtained. A small amount

of metal, in this case tin, was melted at a given temperature above the melting point (superheat) in an evacuated quartz tube, and then held at a determined undercooling temperature so that heat of crystallisation was set free. This heat of crystallisation was caused to operate a stop watch by means of a thermo-couple, galvanometer, and photo cell, and thus automatically to register the time of the beginning of crystallisation.

The results obtained for tin by this method showed that overheating the molten metal reduces, and undercooling increases, the number of crystallisation nuclei. In the case of a homogeneous metal cast into a sand or metal mould, three zones of different crystalline structure are formed: a fine-grained narrow zone at the edge, a broader zone of columnar crystals, and an inner zone of irregularly orientated crystals. The depth, or even the existence, of these zones depends on the casting temperature and the temperature of the mould. For instance, in the case of pure aluminium cast at a temperature of 600 deg. Cent. into a mould at a temperature of 20 deg. Cent., all three zones are clearly formed. As the temperature of the mould is increased, first the fine-grained outer zone and finally the columnar zone disappear. With a casting temperature of 900 deg. Cent. the grain size is much coarser; at the lower mould temperatures only the columnar zone occurs, and as the mould temperature is increased the fine-grained structure appears in the centre and finally displaces the columnar zone entirely.

Scheil explains the origin of the outer and columnar zones as being due to the difference in undercooling as the distance from the edge of the ingot increases. The fine-grained outer zone is formed by relatively strong undercooling; with decreasing undercooling, the formation of nuclei becomes less frequent, so that fully columnar crystals result. He is, however, in some difficulty in explaining the formation of nuclei in the centre. He suggests that some nuclei are produced by the breaking off of the ends of dendrites formed in the outer zones, since mechanical movement of the metal during solidification reduces the grain size and therefore increases the number of nuclei formed. Thus fine-grained structures in pure metal castings can be obtained by avoiding overheating, by vibrating the metal during casting, and by inoculating overheated metal with small pieces of solid metal.

In this connection it is interesting to compare the investigations of L. Northcott, described in a recently published paper.[3] Metal, in this case copper, was poured by the Durville rotary method at, in general, constant superheat into a refractory mould with a water-cooled copper base. An axial section showed that the structure was wholly columnar. Various elements were then added to the copper to the extent of 0·5 to 1·0 per cent., and the resulting ingots were found to consist of columnar crystals at the bottom and equi-axed crystals above. The lengths of the different columnar crystals was the same in any one ingot, and under constant casting conditions their length was taken as a measure of the tendency of the metal to solidify in columnar form. The effect of some of the elements was remarkable. For instance, 1 per cent. of lead in an ingot 2·8in. high produced crystals only 0·25in. long. Northcott allots a "growth restriction factor" to each metal to indicate its relative effect in restricting columnar growth. By tabulating the elements studied in order of their growth restricting effects, and also of several of their physical properties, such as lattice type, atomic volume and radius, crystal structure, interatomic distance, specific heat, and solid solubility, he found that the effect of these properties, if any, is small and is completely outweighed by that of some major property. In attempting to explain the mechanism controlling the growth of crystals from the melt, Northcott describes (1) the effect due to a composition gradient, and (2) the effect due to adsorption. Consider, under the first heading, the case of an alloy which solidifies over a range of temperature. If there is a steep temperature gradient and the second phase diffuses slowly, there may be an enrichment of the lower melting point phase in the liquid layer adjacent to the solid, so that the next layer in the liquid may reach the liquidus temperature of the alloy first, and independent solid nuclei will appear resulting in equi-axial crystals. This, however, although probably a contributory factor, is not the sole cause, for the rates of diffusion of the different elements in solid copper do not follow the order of their growth restriction effects, and, as Northcott has himself shown earlier,[4] continuity in growth may be maintained in copper-rich alloys in spite of several such interferences. He therefore considers that a more complete explanation lies in adsorption.

In the early stages of crystal growth the copper atoms arrange themselves in a lattice to form the crystal and later the foreign atoms become adsorbed on the face of the growing crystal. The greater the degree of adsorption the greater will be the resistance offered to the attachment of more copper atoms to the lattice, so that finally, when they solidify owing to the decrease in temperature, the groups of atoms will not adopt the same orientation as the already solid copper.

There is some similarity between this and the previous explanation in that both are concerned with a screening action, but in the former case it is due to a liquid layer and in the latter to a solid layer. Northcott also shows that there is a relationship between the valency and atomic structure of the added elements and the growth restriction factor, and that the effect of the elements on the crystallisation of copper is a periodic one. The periodicity becomes more pronounced when the elements in the long periods of the periodic table are considered to be arranged in groups of four instead of eight.

Exactly fifty years ago, Roberts-Austen[5] drew attention "to the way in which tenacity and extensibility of metals may be affected by small quantities of metals and metalloids, with a view to showing that the relations between these small quantities of the elements and the masses of metal in which they are hidden are under the control of the law of periodicity." He added to pure gold a number of pure elements to the extent of about 0·2 per cent., and from a measurement of the mechanical properties of the resulting alloys concluded that "metals which diminish their tenacity and extensibility have high atomic volumes, while those which increase these properties have either the same atomic volume as gold or a lower one." The technique of micro-examination had not, of course, advanced very far in those days, but the significance of this careful work in relation to the views on crystallisation now put forward by Northcott appears to call for closer investigation.

REFERENCES.

[1] *Z. Metallkunde*, 1931, 23, 165.

[2] *Z. Metallkunde*, 1937, 29, 404.

[3] *Monthly J.*, Inst. Metals, 1938, 5, 9.

[4] *J.*, Iron and Steel Inst., 1934, 79, 171.

[5] "Phil. Trans.," Roy. Soc., 1888, 179, 339.

The Mechanism of Age Hardening in Copper-Aluminium Alloys.

IT is a commonplace that the more startling advances in industrial metallurgy have been the fruit of empirical discovery rather than planned research, but there are few subjects in which research has been more successful in extending empirical discoveries far beyond their original bounds. The case of age hardening is typical; Wilm's discovery[1] led directly to one alloy or group of alloys, "Duralumin," and there the subject might have rested had not it been recognised that a capacity to age harden is tied up with thermo-dynamic instability and usually with the attempt of a supersaturated alloy to reject a second phase; in the twenty years that have followed the realisation of this principle by Merica, Waltenberg and Scott[2] the number of age-hardening alloys known has increased from one to hundreds, and these new materials include alloys of almost every metal of commerce. Age-hardening properties are to be sought amongst alloys so situated in the equilibrium diagram that they have a homogeneous constitution at one temperature and a duplex constitution at a lower temperature, such as room temperature, *i.e.*, where solid solubility decreases with temperature, as is the case with almost all terminal solid solutions.

It will have been realised from the above paragraph that in the writer's opinion enough was known twenty years ago to allow of prediction being made of the probability of a given alloy being capable of age hardening. Intensive research during these years has led to the discovery of many useful materials, but cannot be said to have yet shown up the inner mechanism of age hardening in such a way that any quantitative treatment of the subject might be made, or the factors leading to unusual hardenability recognised. This is scarcely surprising when the complexity of the subject is considered; for example, variations in degree of purity have startling effects on the rate of hardening of copper-aluminium alloys at room temperature, but little effect at an ageing temperature of 200 deg. Cent., so that not one series of property measurements is needed to fix the characteristics of a given alloy, but tests carried out at a variety of temperatures. Again, different mechanical tests have a habit of giving ageing curves for a single alloy which reach maxima at different times and even more acute lack of parallelism is shown between curves representing changes of other properties such as resistivity, density, lattice parameter, and magnetic susceptibility,

As age hardening was discovered in an aluminium alloy, there has naturally been a tendency to concentrate detailed investigations of age-hardening mechanism on such materials, though extended surveys have been made of beryllium-copper, copper-silver and other age-hardening alloys. It is with the copper-aluminium alloys, of which we have the most complete information, with which we are mainly concerned at present, and particularly the position in that field as recently summarised by Wassermann[3] in a paper which confirms and illustrates in a very beautiful way some of the most recent advances. Wassermann directs attention to three features of the age hardening of copper-aluminium alloys which have recently come to light, namely, the importance of quenching strains, the early changes of microstructure, and the formation of an "intermediate body" during ageing.

When a polished specimen of a 3–6 per cent. copper-aluminium alloy is heated to such a temperature as 530 deg. Cent., so that all the copper is taken into solution, and quenched in ice-cold water, a series of markings develop upon the surface, which run across each crystal in parallel formation, ceasing abruptly at the crystal boundary, as a rule. Such a surface is illustrated in Fig. 1, at a magnification of twelve diameters. If the alloy now be subjected to an ageing treatment for a period rather shorter than that needed to raise the hardness to a maximum, and is then polished and etched, a pattern which is obviously closely related to that shown on the quenched surface, is revealed, as shown in Fig. 2. If the

FIG. 1—*Production of Twin Formation in the Quenching of an Aluminium-Copper Alloy. Section Quenched in Water, Unetched.*

original quenching was less drastic, as, for example, in water at 100 deg. Cent. or oil at 150 deg. Cent., the development of striae, both on the quenched surface and on the aged, polished, and etched material, is much less prominent, but the striae are then accompanied in the aged material by sub-boundary veining, as seen in steels. At the suggestion of Professor Schmid, Wassermann refers to the striae as mechanical twins, whilst Fink and Smith,[4] who discovered the effect in copper-aluminium alloys, describe them as slip-bands. Neither the present paper nor that of Fink and Smith contains any evidence relating to the orientation of the striae relative to the lattice

of the solid solution, but both appear to assume that they run along octahedral planes. Wassermann's illustrations would be reconcilable also with a suggestion, which occurs to the present writer, that the markings represent areas of non-homogeneous deformation, like Lüder's lines in steel; this suggestion again involves the octahedral planes of the aluminium lattice and is rather tentative until the orientation is known. One reason for this suggestion is that Lüder's lines in steel overlie zones which, on tempering at 200 deg. Cent. followed by etching with Fry's reagent are revealed as more deeply etched bands, and the mechanism of Fry's etching has long been regarded as connected with local age hardening and precipitation of iron nitride.[5]

It is not possible at this time to go completely into the history of the type of structure under discussion, but brief reference may be made to the

FIG. 2—*Production of Twin Formation in the Quenching of an Aluminium-Copper-Alloy. The Same Section as Fig.* 1 *Aged Three Hours at* 200 *deg. Cent., Polished and Etched.*

papers of Norbury,[6] who found striae in aged copper silver alloys, and Hansen,[7] who found them in zinc-magnesium alloys. Neither of these authors connected the striae with quenching strains. Even more pronounced development of a "plaid-like" pattern is found in beryllium-copper alloys,[8] both in the supersaturated alpha and decomposing beta-phases, and in zinc-aluminium alloys,[9] in which the effect is closely similar to the formation of martensite from austenite and is due to unavoidable internal strains connected with atomic rearrangement, and not to quenching as such. Somewhat similar in origin are probably the meteorite-like markings in copper-gold alloys near the stage of maximum hardness, shown by Haughton and Payne.[10]

A closely related observation is that of Phillips and Brick[11] that the proportion of compound forming at a previously polished surface of a copper-aluminium alloy during ageing at such temperatures as 400 deg. Cent. exceeds considerably the equilibrium concentration, precipitated $CuAl_2$ either reaching the surface by diffusion or being less soluble in the strained surface. These authors[12] in another paper have also drawn attention to abnormally high values of lattice parameter which result from cooling high purity copper-aluminium alloys faster than an optimum rate, corresponding to air cooling a thin section. As increased copper content raises the lattice parameter, these results would at first sight seem to point

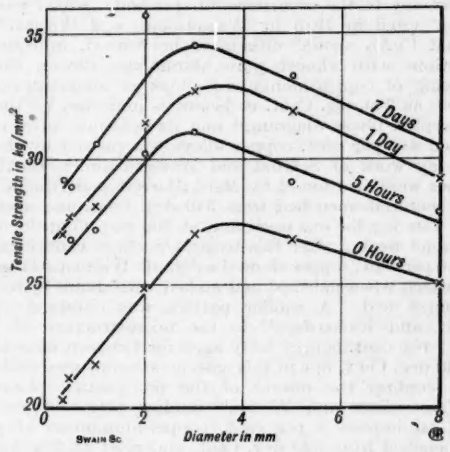


FIG. 3—*Tensile Strength of Aluminium-Copper Wires* (4·8 *per cent. Copper*) *in Relation to the Diameter of the Wire and the Period of Ageing at* 20 *deg. Cent.*

to a decrease in the amount of copper retained in solution as quenching becomes more rapid. Actually, the effect is entirely due to strain, as was shown by the fact that the initial change during ageing the drastically quenched material was towards a larger parameter, whereas in the air-cooled material the parameter fell steadily during ageing. The effect of the quenching strain on the subsequent rate of ageing may be gauged from Fig. 3, which shows the progress of ageing at room temperature of 4·8 per cent. copper-aluminium wires of various diameters quenched in ice water from 525 deg. Cent. in terms of change of breaking load. The highest values of strength correspond to a diameter of about 3 mm.

It may be remarked that Wassermann does not bring forward any evidence that the development of striae is more than a localised effect near the surface or that it produces any radical changes in the course of hardening, a point on which Fink and Smith's hardness, tensile strength, and other results were unconvincing. Further work seems to be needed before it can be accepted that the strained regions

make any considerable contribution to the hardening of the specimen as a whole, and are the origin of double ageing peaks, as Fink and Smith suggest.[4] It is not unlikely, however, that the absence of quenching strains in material cooled by oil quenching or quenching in boiling water is responsible for their resistance to corrosion differing from that of material quenched in cold water. The distribution of strains is in accordance with the observation of Shinoda and Tatibana,[13] who applied the Schlieren method to the study of the quenching of Duralumin, that heat leaves the material in a different way during water quenching from that occurring during oil quenching; the heat is extracted mainly by convection during oil quenching, but mainly by flow normal to the surface in water quenching.

The remainder of Wassermann's paper deals with the formation in age-hardened copper-aluminium alloys of an "intermediate body"—a half-way stage in the rejection of $CuAl_2$ from the original solid solution. Great importance is attached in Germany to the occurrence of this body, which was first noted in 1935 by Wassermann and Weerts.[14] That $CuAl_2$ should ultimately be formed, in equilibrium with almost pure aluminium, during the ageing of copper-aluminium alloys at temperatures such as 200 deg. Cent. or below, is indicated by the accepted phase diagram,[15] and its presence in over-aged 4·5 per cent. copper alloys was proved by the X-ray work of Schmid and Wassermann.[16] $CuAl_2$ lines were also found by Mehl, Barrett and Rhines[17] in material quenched from 540 deg. Cent. and aged at 250 deg. for one week and at 300 deg. Cent. for a second week; after this treatment alloys containing 1–6 per cent. copper showed a definite Widmanstätten pattern when polished and etched with dilute hydrofluoric acid. A similar pattern was obtained by Dix and Richardson[18] in the microstructure of a 0·7 per cent. copper alloy aged for fourteen days at 200 deg. Cent. but in this case no attempt was made to confirm the nature of the precipitated phase. Wassermann and Weerts[14] studied single crystals of an impure 5 per cent. copper-aluminium alloy quenched from 520 deg. Cent. and aged at 200 deg. and 300 deg. Cent.; precision X-ray work indicated that whereas $CuAl_2$ is rejected in its normal crystallographic habit at 300 deg., exposure at 200 deg. for twenty-four hours results in the formation of a different body, which is, however, closely related to $CuAl_2$. The intermediate body crystallises in tetragonal cells having the axes 8·2 A, 11·6 A, and probably containing 16 copper and 32 aluminium atoms. The tetragonal axes of the intermediate body were found to be parallel to cube axes of the aluminium lattice; the arrangement resembles closely that found by Mehl, Barrett and Rhines[17] for the orientation of $CuAl_2$ in aluminium, where the $CuAl_2$ plates lie in cubic planes of the aluminium. Exposure of the 5 per cent. alloy for twenty-four hours at 200 deg. Cent. was stated clearly to develop the intermediate body.

In the present paper, Wassermann traces the formation of the intermediate body and of $CuAl_2$ in single crystals of two alloys containing 4·8 and 5 per cent. copper, made up from the purest aluminium and electrolytic copper quenched from 530 deg. Cent. Ageing at 150 deg. Cent. results in the appearance of traces of the intermediate body on the X-ray photograms after thirty days; at 200 deg. the first signs of the intermediate body are seen after nineteen hours, and those of $CuAl_2$ after seventy days—this observation appears to contrast with Dix and Richardson micrographic work cited above. At 300 deg. the situation is as shown in Fig. 4; the intermediate state is strongly developed in one hour and afterwards gradually vanishes, whilst the formation of $CuAl_2$ is well advanced in fifty hours and complete in 1000 hours.

It is difficult exactly to compare these times and temperatures with those required for the production of maximally hardened material, since hardness results are not readily available for single crystals but the intermediate body is probably formed before the stage of maximum hardness. The orientation of the intermediate body differs from that assumed for the strain markings, but it is possible that initial change occurs within the strain bands, which are usually at least 200 atomic diameters in width. The intermediate body formed in these alloys may be

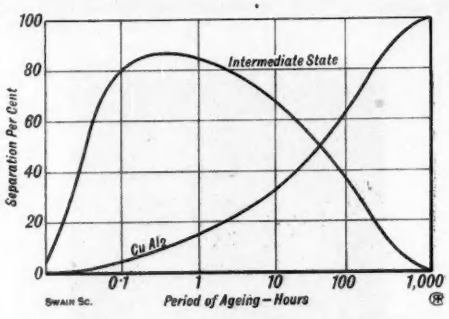


FIG. 4—*Separation and Transformation, Intermediate State/$CuAl_2$ in relation to the Period of Ageing at 300 deg. Cent.*

typical of many intermediate bodies produced in alloys which tend towards an equilibrium involving rejection of a second phase. Perhaps the closest analogy is Fe_3C, which is rejected from steels at normal rates of cooling when the equilibrium product would be graphite; another example is the phase, probably Al_3Mg_2 rejected by aluminium-magnesium silicide alloys instead of magnesium silicide, during low-temperature ageing, as shown by Mehl, Barrett and Rhines. The explanation of such initial metastable precipitations is undoubtedly that the diffusion mechanism at work in the alloy does not allow of the collection together of atoms to form the final equilibrium product in the necessary proportions, and, at the same time, in an arrangement convertible into the final form by a simple redistribution, as by shear or twinning. This must often be the case when the phase tending to form has a somewhat complex lattice structure, as have $CuAl_2$ and graphite; another cause is that the formation of a nucleus containing only stranger atoms, *e.g.*, Mg_2Si, in a solvent metal, *e.g.*, aluminium, has a low probability. The intermediate body is probably a structure generated by a homogeneous reaction from a concentrated area in the solution, and just as in martensite forming from austenite the new crystals appearing fully developed and only increasing later in number. These partially converted areas are later presumed to react interfacially with the adjacent solid solution lattice and form the final equilibrium product, which then builds up as the reaction continues by diffusion through the first formed compound layer. The final precipitate first appears as small plates in large numbers, and later are built up to some extent in area, but mainly in thickness.

Only in a limited proportion of age-hardening alloys is the mechanism sketched above likely to apply, for in many cases the phase which is endeavouring to separate is simple in lattice type (it may even be a metal, as in copper-silver alloys), so that it is not along these lines that age hardening is likely to be explained quantitatively. Such studies are, however, invaluable to the elucidation of the general problem of reactions in alloys.

REFERENCES.

[1] Wilm, *Metallurgie*, 1911, 8, 225.
[2] Merica, Waltenberg, and Scott, U.S. Bur. Stnds., Sci. Pap. 347, 1919.
[3] Wassermann, *Zeit. f. Metallkunde*, 1938, February, 30, 66.
[4] Fink and Smith, Amer. Inst. Min. Met. Eng., Tech. Pub. 865, 1937.
[5] Desch, "Metallography," 1937, 338.
[6] Norbury, *Journ.* Inst. Metals, 1928, 39, 145.
[7] Hansen, *Journ.* Inst. Metals, 1927, 38, 216.
[8] Masing and Dahl, "Beryllium and its Alloys," 1932, 180.
[9] Hanson and Gayler, *Journ.* Inst. Metals, 1922, 27, 267.
[10] Haughton and Payne, *Journ.* Inst. Metals, 1931, 46, 457.
[11] Phillips and Brick, *Metals and Alloys*, 1934, 5, 204.
[12] Phillips and Brick, Amer. Inst. Min. Met. Eng., 1934, 111, 94.
[13] Shinoda and Tatibana, Tr. Soc. Mech. Eng., Japan, 1937, 3, 151.
[14] Wassermann and Weerts, *Metallwirtschaft*, 1935, 14, 605.
[15] Hansen, "Der Aubfau der Zweistofflegierungen."
[16] Schmid and Wassermann, *Metallwirtschaft*, 1928, 7, 1334.
[17] Mehl, Barrett, and Rhines, Amer. Inst. Min. Met. Eng., 1932, 99, 203.
[18] Dix and Richardson, Amer. Inst. Min. Met. Eng., 1926, 73, 560.

Austenitic Chromium-Manganese Steels.

THE chromium-manganese steels have been the subject of a considerable amount of investigation, especially directed to their use for corrosion- and heat-resisting purposes.[1] Much attention has been given to them in a review of the new corrosion- and heat-resisting steels by H. Hougardy and G. Riedrich,[2] and they form the subject of a recent article by J. H. G. Monypenny.[3] The general conclusion is that the most suitable variety of chromium-manganese steel for both acid and heat resistance is the austenitic type. Hougardy and Riedrich give steels 1 to 3 mentioned in Table I as typical of the austenitic chromium-manganese acid-resisting steels and steel 4 as a heat-resisting steel. The first-mentioned steel (Table I) is a completely austenitic steel, but in the absence of molybdenum and nickel the chromium content is not sufficient to secure good acid resistance. The addition of 1 per cent. of nickel and 1 per cent. of molybdenum to this steel had the effect of greatly reducing the loss of weight in weak acids (Table II). Steel of this general type has been introduced for domestic purposes under the name of "Roneusil," a material of the colour of silver, and stated by K. Fuchs[4] to be resistant to food, fruit, and organic acids, but not to the stronger mineral acids. The second steel is not fully austenitic, but consists of ferrite and austenite. Photomicrographs are given by Hougardy and Riedrich to illustrate the proportions of these two constituents, which they summarise as follows :—

TABLE I.—*Austenitic Chromium-manganese Acid- and Heat-Resisting Steels.*

Steel No.	Composition, per cent.					Other elements.
	C.	Mn.	Cr.	Mo.	Ni.	
1	0·10	18	9	1	1	—
2	0·10	9	18	—	1	—
3	0·10	9	16	—	2	Ti, Ta, Nb
4	0·10	18	10	—	—	Si4, Ti, Ta, Nb

Steel No.	Yield point, tons per sq. in.	Ultimate stress, tons per sq. in.	Elongation, per cent. ($l = 5\,d$).
1	19—28	47·5—54	>35
2	19—25·5	44·5—51	>35
3	22—28	47·5—54	>35
4	22—25·5	47·5—54	>50

TABLE II.—*Effect of Nickel and Molybdenum Additions on the Corrosion Resistance of* 18 : 9 *Manganese-chromium Steel.*

Steel No.	Composition, per cent.					
	C.	Si.	Mn.	Cr.	Ni.	Mo.
1	0·12	0·76	16·77	9·88	0·58	—
2	0·11	0·77	16·78	9·92	1·48	0·76

Loss of Weight of Polished Specimens at 20 *Deg. Cent. in Grammes Per Square Metre Per Hour.*

Corroding medium.	Steel 1.	Steel 2.
10 per cent. acetic acid	0·2570	0·0523
20 per cent. tartaric acid...	0·3898	0·0602
20 per cent. citric acid	0·0635	0·0035

Mn.	Cr.	Ni.	Austenite.	Ferrite.
9	16	2·5	100	Nil
9	16·5	2·0	85	15
9	17·5	1·0	70	30
9	20	1·0	30	70

The elongation falls as the amount of the ferrite constituent increases, and when the chromium content is high, *i.e.*, in the vicinity of 15 per cent., the presence of nickel is always necessary to preserve the austenitic condition and maintain good properties. According to M. Broich,[5] chromium-manganese steel of the 18 : 9 type has been used, in the form of tubes, for well filters, &c., under the name of "Remanit." Titanium, tantalum, or niobium is added to the third steel (Table I) to prevent "weld decay" to which the plain chromium-manganese steel is subject. The fourth steel (Table I) is a heat-resisting steel, and the titanium is added to this in order to maintain a good impact figure on prolonged heating. The influence of the titanium in maintaining the impact figure in the reheated steel is illustrated in Fig. 1. This steel, in spite of its high silicon content, is fully austenitic and therefore non-magnetic. It is stated to be almost identical with 25 : 20 chromium-nickel steel in electrical resistance and thermal conductivity and to have a distinctly higher creep limit at 600 deg. to 750 deg. Cent. As well as summarising the structure and properties of the steels, Monypenny[3] criticises the diagram put forward by Schmidt and Legat[4] on the ground, among others, that the austenitic region is too extensive. This criticism has almost simultaneously been recognised and, in part, met by H.

Legat,[6] who has continued his investigations of the chromium-manganese steels, mainly with a view to determining the compositions suitable for use as non-magnetic materials of construction. The field which he desired to cover was that of stable austenite, and the revised limits of composition of this field are indicated by Legat in the diagram (Fig. 2), which represents the region 0 to 20 per cent. chromium and

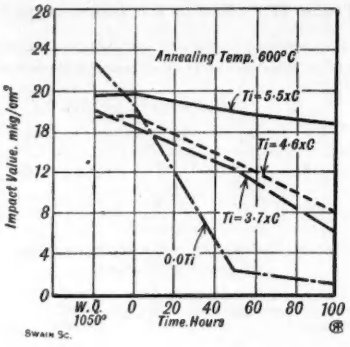


FIG. 1—*Influence of Titanium on the Impact Value of Silicon-manganese-chromium Steels (C. 0·10 to 0·16; Mn. 16·6 to 19·6; Cr. 7·5 to 8·3; Si. 3·6 to 4·3 and Ti. nil to 0·57 Per Cent.), Annealed at 600 Deg. Cent. After Water Quenching from 1050 Deg. Cent.*

10 to 44 per cent. manganese, near the iron corner of the iron-chromium-manganese constitutional diagram. The boundary of the austenite region which was given in the previous paper is now shown to extend to too high a chromium content and too low a content of manganese. It corresponded approximately to the line $B'_1 B_1 C_1$, which now represents the limits of composition of steels which are wholly austenitic in the air-cooled condition, while the line $B'_2 B_2 C_2$ gives the limiting compositions of steels which remain austenitic after all types of heat treatment. The structure of the alloys corresponding to points above the line $B'_1 B_1 C_1$ consisted of ferrite grains in an austenitic ground mass, below $B'_2 B_2 C_2$ the alloys contained austenite grains only. With increasing carbon content the austenite field at 1100 deg. Cent. is extended, but, on cooling to lower temperatures, the austenite is no longer stable, but decomposes with precipitation of carbide.

The region was explored by Legat by the aid of fifty-six melts of 2 kilos. to 5 kilos. made in the high-frequency furnace. The steels were cast and forged into bars, the finishing temperature being about 850 deg. Cent. All the low-carbon steels up to 15 per cent. chromium and 43 per cent. manganese were forgeable; one with 53·7 per cent. of manganese was not. To some of the 14 : 16 chromium-manganese steels additions were made of silicon up to 1 per cent., copper up to 2·2, nickel up to 4·1, and molybdenum

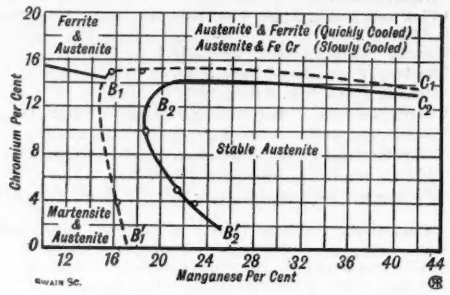


FIG. 2—*Constitutional Diagram of the Chromium-manganese Steels (Containing 0·1 to 0·2 Per Cent. Carbon) at Room Temperature.*

up to 5·3 per cent. All these were forgeable, though when the copper exceeded 4 per cent. they were reported as completely unforgeable.

For mechanical testing, tensile and notched-bar test pieces were obtained from the 14·4 mm. diameter forged bar. Tests were made on the steels as forged, after reheating the forged material to 400 deg. and to

TABLE III.—*Mechanical Properties of the Forged Steels.**

Chromium, per cent.	0·5.			4·0.					
Manganese, per cent.	18	31	38	16	22	23	31	35	43
Yield point, tons per sq. in.	39	37	41	46	38	38	39	39	37
Ultimate stress, tons per sq. in.	61	51	51	64	56	52	50	48	46
Elongation, per cent.	42	41	36	16	38	40	40	38	30
Reduction of area, per cent.	52	60	62	20	56	56	56	56	56
Chromium, per cent.	8·0								
Manganese, per cent.	13	18	22	27	30	33	36	38	40
Yield point, tons per sq. in.	41	38	33	39	44	49	48	43	29
Ultimate stress, tons per sq. in.	75	60	56	51	51	51	51	48	39
Elongation, per cent.	20	36	50	40	32	22	36	38	42
Reduction of area, per cent.	22	50	58	58	60	62	60	58	62

* Kgm. per sq. mm. converted to the nearest ton per sq. in.

600 deg. Cent., and also after quenching from 960 deg. and from 1100 deg. Cent. For comparative purposes four commercial steels were also tested, viz., 13 per cent. manganese steel, 18 : 8 chromium-nickel steel, and high- and low-carbon 8 : 36 chromium-manganese steels.

The results of these tests are given in a series of curves, the indications of which may be considered in conjunction with the diagram (Fig. 2) by the aid of typical results read from the curves (Table III). On

passing across the diagram (which, in the original paper, is marked out in contour lines of tensile strength and yield ratio) from left to right at different chromium contents, there is a fall in tensile strength with increase of manganese content until a certain manganese, depending on the chromium present, is reached beyond which tensile strength, elongation and reduction of area change only very slightly. Thus, with 4 per cent. of chromium the properties are very similar in steels covering a range of manganese from 22 to 43 per cent. Similarly, in traversing the diagram vertically at a fixed manganese content the properties remain constant until a certain chromium content is reached, *e.g.*, about 14 per cent. of chromium with 16 per cent. of manganese, or 10 per cent. of chromium with 38 per cent. of manganese, when an increase of hardness occurs, accompanied by a fall in impact figure. The chromium limit of the austenitic region falls slightly with increase of manganese content. The mechanical properties of the 14 : 16 chromium-manganese steel in the forged condition are not much affected by the addition of copper or of molybdenum up to 2·5 per cent. The steel is considerably hardened by 5·5 per cent. of molybdenum and softened by 4 per cent. of nickel (Table IV),

Table IV.—*Effect of Additions to* 14 : 16 *Chromium-manganese Steel on the Mechanical Properties in the Forged Condition.*

Addition.	Nil.	4% Ni.	2% Cu.	2·3% Mo.	5·5% Mo.
Yield point, tons per sq. in.	44	39	46	44	58
Ultimate stress, tons per sq. in.	60	54	61	60	69
Elongation, per cent. ...	42	38	35	40	18
Reduction of area, per cent.	60	60	58	62	20
Impact figure Mkg./cm.[2]	12	12·5	10	12·5	3·5

which produces a more stable austenite and also gives rise to the most readily workable steels of the series.

The tests referred to above were all made on forged bar, which must to some extent have been affected by cold work. This would account for the very high yield ratio observed, reaching its maximum in the proximity of the composition chromium 7, manganese 33 per cent. The diagrams include a certain number of tests on material as quenched from 1100 deg. Cent. These are noted in Table V.

Table V.—*Mechanical Properties of the Steels Water Quenched from* 1100 *Deg. Cent.*

Carbon, per cent. ...	0·24	0·35	0·11	0·14	0·20	0·19
Chromium, per cent.	15·12	15·04	8·12	14·0	14·0	14·0
Manganese, per cent.	15·72	15·68	33·12	16·0	16·0	16·0
Other elements, per cent.	—	—	—	Ni. 2·65	Cu. 0·94	Mo. 2·28
Yield point, tons per sq. in.	25	27	19	22	24	25
Ultimate stress, tons per sq. in.	52	50	32	42	46	50
Elongation, per cent.	55	50	72	60	60	58
Reduction of area...	65	55	62	68	66	64

Machinability tests were made on all the experimental steels, the best results being obtained with 4 per cent. chromium and over 20 per cent. manganese, 8 per cent. chromium and 20 to 40 per cent. manganese, and with the 14 : 16 chromium-manganese steel containing 1 to 4 per cent. of nickel. Increase in carbon content had a pronounced effect in increasing difficulty of machining, the effect becoming evident at 0·2 per cent. carbon in 15 : 16 chromium-manganese, and at 0·1 per cent. carbon in 8 : 38 chromium-manganese. The difficulty increased to such an extent that the steels with over 0·3 per cent. of carbon could not be economically machined. This was confirmed on the two commercial 8 : 36 chromium-manganese steels with 0·15 and 0·59 per cent. of carbon respectively.

The chromium-manganese steels offer a wide field for development. Their possibilities and also their limitations have been realised more fully during the last few years. It seems unlikely that they will equal the chromium-nickel steels in any respect except resistance to sulphurous gases at high temperatures, but for a large number of corrosion resisting purposes they form an economical alternative. The studies which have been made of their constitution are of value in the selection of suitable compositions, but the effect of additions of other elements needs further exploration, since copper, molybdenum, and especially nickel, have shown such beneficial results.

References.

[1] See The Metallurgist, February 26th, 1937, page 11.

[2] *Metallwirtschaft*, December 17th, 1937, 16, 1329.

[3] *Metallurgia*, November, 1937, and January, 1938.

[4] *Metallwirtschaft*, 1937, 16, 215.

[5] *Internationale Röhrenindustrie*, 1937, 2, 3.

[6] *Archiv für das Eisenhüttenwesen*, January, 1938, 1937/38, 11, 337.

Grain Boundaries.

A paper which may be the prelude to much interesting work has been communicated to the American Society for Metals.* It deals with the "Solubility of Copper in the Grain-Boundary Material of a Solid Solution of Copper in Zinc." The opinion has often been expressed that in a pure solid solution the solubility of one metal in another is different in the intercrystalline boundaries and in the crystals themselves. In order to subject this hypothesis, as applied to a weak solution of copper in zinc (the eta phase) to an experimental test, G. R. Dean and Wheeler P. Davey have devised a special technique for sampling and analysing material derived from the grain boundaries and from the interior of the grains. This consists of exposing a very restricted area of a specimen covered with bakelite lacquer by cutting through the bakelite film at the grain boundaries or other selected region on the specimen. Solution from this region was effected by electrolytic means, about 10 milligrammes being removed. The copper and zinc dissolved were deposited on a platinum electrode, redissolved, the solution evaporated on the positive carbon of the arc, and the material so obtained was subjected to a quantitative spectrographic analysis in direct comparison with specially prepared standards. It was claimed that specimens of the order of 0·5 milligramme of solid solution containing 1 to 2 per cent. of copper could be analysed quickly by the spectrograph with an accuracy of about 10 per cent.

Precautions were taken with respect to purity of materials, preliminary annealing and the avoidance of errors due to preferential solution of zinc or to external variations in spectrographic technique.

* G. R. Dean and Wheeler P. Davey, "Trans.," Am. Soc. for Metals, 1938, 26, 267.

Typical values of copper content were as follows :—

Chemical analysis of massive sample. Per cent.	Spectrographic analysis. Body of grain.* Per cent.	Grain boundary. Per cent.
1·07	1·1	0·8
1·65	1·6	1·4
2·07	2·1	1·8

* Closely adjacent to the grain boundaries.

The differences in composition indicated above are so striking and so large in comparison with the errors of measurement that it is hard to escape the conclusion that the copper content of the solid solution is much lower in the grain boundaries than in the body of the crystals. The possibility that the irregular distribution of copper was due to inadequate annealing was guarded against by making the cuts in the lacquer, by which the grain body was sampled, close to and parallel with the grain boundaries. As shown by the above figures, the spectrographic analysis of samples near the boundaries gave results in close agreement with the overall analysis of the specimen.

The authors indicate how their results may be interpreted in the light of the formation of a Gibbs layer at the interface, the decrease in copper content implying that the addition of copper tends to raise the interfacial tension between zinc crystals, an effect which would be likely to occur if the surface tension of copper were higher than that of zinc at the annealing temperature (410 deg. Cent.). It is known that the surface tension of copper is 50 per cent. greater than that of zinc when measured at the respective melting points of the metals. There is therefore some little justification for supposing that it may be greater at the annealing temperature also, and that a Gibbs layer may be formed. If this is so, the copper-rich alloy (the alpha phase) should also show an impoverishment of copper, *i.e.*, an excess concentration of zinc, in the grain boundaries. This conclusion calls for early examination and in other directions also there is probably a considerable field for the application of micro-analytical technique to problems of this kind.

Lead-Calcium Alloys for Accumulators.

IN two February journals, viz., *Metals and Alloys* and *Zeitschrift für Metallkunde*, articles have appeared indicating the advantages possessed by lead-calcium alloys over the usual lead-antimony alloys for accumulators or storage batteries. The American article[1] takes the form of an extended abstract of the papers which have been published during the last two or three years in the "Transactions" of the Electrochemical Society of America ; the German article[2] is a record of tests and experience published by the laboratories of the Accumulatorenwerk Hoppecke of Zoellner and Co.

In both articles reference is made to some of the disadvantages of the use of the antimony-lead alloy which it is sought to replace. The production of a serviceable lead accumulator requires the employment of a material which is sufficiently hard to withstand deformation produced by external forces, vibration, shock, &c., or by volume changes in the electrodes. At the same time, it must resist anodic and cathodic corrosion in sulphuric acid. Expansion, especially of the positive grids, is likely to take place under stress set up by the corrosion products if the material is too soft. Soft lead does not meet the mechanical requirements. The alloy with 5 to 8 per cent. of antimony is better fitted to do this, but suffers from other defects. Owing to anodic corrosion of the antimonial lead positive grids, antimony is dissolved from the antimony-rich eutectiferous regions, thus weakening and ultimately disintegrating the grid. The dissolved antimony passes through the acid to the negative plate, where it is deposited, forming local galvanic couples which increase the rate of self-discharge. Sulphation, which is relatively slow for pure lead, is greatly accelerated by the presence of antimony. The paper of Hoehne describes the results achieved in Germany by the adoption of the American suggestion for replacing lead-antimony by lead-calcium which is electropositive to lead and has adequate mechanical properties.

The alloys recommended for use in America contain less than 0·1 per cent. of calcium. This amount is soluble in lead at the peritectic temperature 328 deg. Cent. and the solubility falls to 0·04 per cent. at about 260 deg. Cent. and to 0·01 per cent. at room temperature. The 0·1 per cent. calcium-lead alloy therefore consists of α-solid solution of calcium in lead with a precipitated phase Pb_3Ca.

The diagram of Schumacher and Bouton is reproduced below. The mechanical properties were determined by Schumacher and Phipps and are reviewed by J. Z. Briggs. The maximum tensile strength of 8000 lb. per square inch obtained in the age-hardening 0·1 per cent. calcium alloy is similar to that of a 9 per cent. antimony-lead alloy. There is no danger of softening on storage, as there was no decrease in the tensile strength after eight years. Cold rolling increases the tensile strength of 0·04 per cent. calcium-lead from 5200 lb. to 6600 lb. per square inch, but when the calcium exceeds 0·08 per cent. the strength of the alloy is slightly reduced by cold rolling. Reassurance as to the mechanical properties of calcium-lead alloys is also contained in a paper by K. von Hanffstengel and H. Hanemann[3] on "The Alloys of Lead with Calcium and Lithium." The hardness of the calcium-lead alloy is stated to be equal to that of the antimony-lead alloy if it is water-cooled after casting and aged. The Brinell hardness values are stated to be as follows for an alloy with 0·1 per cent. calcium :—

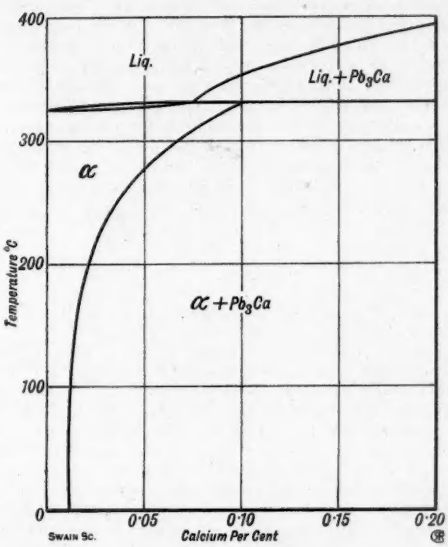
Pressed at 270 deg. Cent.	5·6
The same, then quenched from 310 deg. Cent.	9·6
Cast in warm mould, air cooled	9·0
The same, annealed at 290 deg. Cent. and quenched in water	10·9
Cast, released from mould and quenched	11 to 13

Incidentally, it may be noted that equally high hardness was obtained in 0·07 per cent. lithium-lead alloys. When calcium 0·1 and lithium 0·07 per cent. were both present the hardness reached 18·5, but the use of this alloy in accumulators would be quite impracticable on account of its susceptibility to corrosion. The same authors report on the fatigue properties[4] of the 0·07 per cent. calcium alloy (hardness 8·4) and find a value of 1·0 kilo. per square millimetre for 10 million alternations or about three times the value for pure lead. An antimonial lead of hardness 10·8 gave 1·25 kilos. per square millimetre.

The calcium-lead alloy used by Hoehne contained sodium in addition. It was made by melting together

1 part of lead alloy, containing 2·7 to 2·9 per cent. of calcium and 0·45 per cent. of sodium, and 24 parts of soft lead. An appliance for use in making up these hardener alloys is described in v. Hanffstengel and Hanemann's paper.[3]

The maximum sodium content of Hoehne's alloy probably did not amount to 0·02 per cent. and no detrimental effect on resistance to corrosion is reported. The alloy is referred to by him as a calcium alloy. The alloy which he designates "hard lead" contained 5 to 8 per cent. of antimony. The tests described have extended over eight or twelve months and are not yet finished; 135 accumulators of different types were fitted with calcium-lead plates, 73 were run under practical conditions, and the rest in the laboratory. The mechanical properties of the calcium-lead were quite equal to those of the antimony-lead and, moreover, remained constant. Positive and negative grids containing 0·08 to 0·09 per cent. calcium after eight to twelve months' service gave Brinell hardness values of 13 to 17. The demand for mechanical strength is not so great in the negative as in the positive grids, and as negative plates the calcium alloy showed exceedingly good service, and behaved quite as well as "hard lead." In the positive grids very considerable difficulties were encountered, originating not in the behaviour of the alloy itself, but in the method of jointing the plates to connecting bars. It was found, however, that a junction of antimonial-lead and calcium-lead was more satisfactory than one made between two calcium-lead



Calcium-Lead Diagram.[5]

components. No marked corrosion had occurred at the junction of calcium-lead and antimonial-lead after one year's service, and it was found that anodic corrosion of soldered or burnt-on joints could be repressed if only one of the two parts joined together was made of calcium-lead.

The advantage of the electropositive character of the calcium was fully confirmed. The charging voltage could be higher without gassing and this advantage increased with the number of chargings and dischargings. The charging voltage of the calcium-lead cell remained constant while that of the antimony-lead cell fell with the increasing number of cycles of charge and discharge. The dependence of charging conditions on the nature of the electrodes is shown in Table I. The cells used had a capacity of 12·6 ampere-hours at 1·26 amperes discharge current. The low value for hard lead is due to the dissolved antimony which is deposited on the negative plate, and the fall in charging voltage with increased number of discharges is due to the continuous progress of this effect. Calcium similarly goes into solution at the positive grid, but, being electropositive to lead, it is not deposited on the negative plate. The antimony dissolved from the positive grid and deposited on the negative plate has another disadvantage in that it

TABLE I.—*Charging Voltage in Relation to the Nature of the Electrodes.*

Cell No.	Nature of the		Charging voltage at 2·8 amperes charging current.
	Positive.	Negative.	
1	Hard lead ...	Hard lead ...	2·52
2	Hard lead	Calcium-lead ...	2·51
3	Calcium-lead ...	Hard lead	2·71
4	Calcium-lead ...	Calcium-lead ...	2·73
5	Soft lead	Hard lead	2·72
6	Soft lead	Calcium-lead ...	2·78

increases the rate of self-discharge of the battery. Tests were made on the same six cells to find out how calcium-lead plates behave in relation to self-discharge. The six cells after seventy cycles of charging and discharging were left in the charged state for forty-six days and their capacity determined. After a further seventy-two cycles, making 142 in all, they were left in the charged condition for thirty-three days. Table II shows the loss of capacity of the cells expressed as a percentage of the capacity before rest. The capacity tests were made with a discharge current of 1·26 amperes; they were stopped at a voltage of 1·8. As far as could be

TABLE II.—*Self-discharge of Cells in Relation to the Nature of the Electrodes.*

Cell No.	Loss of capacity during		Loss of capacity of the	
	First rest (46 days).	Second rest (33 days).	Positive.	Negative.
	Per cent.	Per cent.	Per cent.	Per cent.
1	74	56	54	50
2	64	64	58	61
3	18	12	—	13
4	12	11	—	13
5	7	9	4	11
6	8	5	5	—

ascertained, the capacity loss of the separate electrodes is given for the second rest period. The determining factor of the order of magnitude of the self-discharge is the nature of the positive plate. The self-discharge of cells with calcium-lead positives amounts to only one-fifth of that of cells with antimonial-lead positives.

The American paper is a brief, but useful, guide to the original work on lead-calcium alloys for storage batteries. There is not a great amount of novel detail in the German paper, but it is interesting as a record of extensive trials, which, on the whole, confirm the value of the American experiments. Hoehne hints at certain difficulties which may be encountered in developing the alloys for use in accumulators. Some of these, associated with the welding of the

alloys, are dealt with in detail; others, judging from experience with lead-calcium bearing metals, may be casting difficulties, such as segregation and loss of calcium on melting. The density of calcium-lead is 11·33, as compared with 10·95 for antimony-lead. The calcium-lead is therefore 3·6 per cent. heavier, and this increase in lead consumption would have to be taken into account by the industry, possibly necessitating re-design of various parts to reduce weight. Hoehne points out that the substitution of lead-calcium for lead-antimony represents a revolutionary innovation for the accumulator industry. It can only be brought about gradually at a rate limited by the need of gaining practical experience of the new alloy. There can be no question of complete replacement of antimony-lead by calcium-lead at an early date. Nevertheless, Hoehne advocates a determined effort to overcome existing defects, not only because calcium-lead is proved to be more efficient electrically as a material for grids, but also because the replacement of antimony by calcium would constitute a valuable contribution to national economy.

REFERENCES.

[1] Janet Z. Briggs, "Lead-calcium Alloys for Storage Batteries," *Metals and Alloys*, Feb., 1938, 9, 49.

[2] Erich Hoehne, "On the Behaviour of Lead-calcium-sodium Alloys in Accumulators," *Zeit. Metallkunde*, Feb., 1938, 30, 52.

[3] *Zeit. Metallkunde*, 1938, 30, 50.

[4] *Zeit Metallkunde*, 1938, 30, 51.

[5] *The lower limit of the Liq.+α region extends to* 0·1 *per cent.*

Books and Publications.

Principles of Powder Metallurgy. By W. D. JONES. Demy 8vo, pp. xii+199. London: Edward Arnold and Co. 1937. Price 15s.

NOTWITHSTANDING the fact that the technique of powder metallurgy received probably its first public notice as long ago as 1828, when Wollaston described the production of massive platinum from powder by sintering, the process until recently appears to have excited little or no interest excepting from an academic standpoint (as exemplified, for instance, by the work of Spring and Masing), and as a convenient method for working up compact masses of refractory base metals, such as tungsten, which experience had proved could be manipulated by fusion and casting only with the greatest of difficulty.

The realisation that the techniques of sintering and fritting not only offer alternatives to the traditional procedure involving casting and mechanical working, but also possess entirely new potentialities, can hardly yet be said to be at all general. The mechanical properties of a metal employed in the cast state may be varied by controlling casting conditions and by subsequent heat treatment. Where the metal is subjected to mechanical deformation such as rolling, further control of the final properties is possible. Powder metallurgy offers even more extensive possibilities, since the physical and chemical phenomena which so largely govern the initial condition of solidified metals and alloys may apparently be regulated to a degree quite impossible by any other method.

It would be idle to suppose that any epoch-making advances can be made until considerably more is known of the physics of metal powders and of the fundamental principles of the sintering and fritting processes. As the author points out in his preface, a wide literature on these subjects does exist, but it is sadly scattered and mainly in foreign tongues. Conflicting hypotheses and theories, and much unsound observation, do nothing to clarify the position. Dr. Jones is to be complimented on the able manner in which he has set out to introduce order into this chaos, and on presenting the first English treatise upon what is virtually a new branch of metallurgy.

Perhaps because the subject has hitherto attracted such limited attention, it is to be expected that some difficulties should arise in devising a systematic nomenclature. Tentative suggestions in this connection have previously been made by Skaupy, Sauerwald, and others. Whilst technologists may quarrel with the author's differentiation between the terms "fritting" and "sintering," few will be unwilling to concede the necessity for such a distinction. The word "compact," too—a new word in the science of metallurgy—appears difficult to replace by any more appropriate name. Without venturing far into the controversy which still surrounds the causes of the abnormally high hardness values attainable in certain sintered metallic masses and ignoring innumerable expedients which are adapted either to allow or avoid porosity, it may be claimed with some certainty that powder metallurgy has vested both properties with a new interest. Of the hardness values possible (180 Brinell in the case of copper, according to Trzebiatowski) little use appears so far to have been made, but of porosity, controlled both as to quantity and quality, advantage has already been taken on porous bearings, and there are indications that its significance as a means of controlling the purely physical and mechanical properties of a wide group of metals has not been overlooked.

Dr. Jones's book should do much to stimulate research in this country in what is undoubtedly destined to become a highly important branch of metallurgy.

Metallography. By C. H. DESCH. Fourth edition (revised and enlarged). Demy 8vo, pp. viii+402. 1937. London: Longmans, Green and Co., Ltd. 21s.

DR. DESCH'S "Metallography" is so well known and universally appreciated that a very brief notice will suffice. This new edition will be welcomed by all serious students of metallography. While the same general plan is retained, Dr. Desch with the lucid, presentation which characterised the earlier editions, has brought the subject matter up to date by the inclusion of the very latest information about the thermal and microscopical study of metals and alloys, and the addition of an account of the new knowledge gained by electron diffraction and X-ray methods. The scientific metallurgist will find the book invaluable, not only in its theoretical, but in its practical treatment of the subject, and any user of metals who reads the book will not only obtain a well-balanced view of the principles of metallography, but will find in many of the more general sections (like those on "The Crystallisation of Metals and Alloys," "The Growth of Metallic Crystals," "Corrosion," "Mechanical Deformation," &c.) the scientific background for many of his practical problems.